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Experiments on Brittle Fracture of Steel Plates

D. K. FELBECK and E. OROWAN

Technical Report No. 1
Office of Naval Research
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EXPERIMENTS ON BRITTLE FRACTURE OF STEEL PLATES

TECHNICAL REPORT NO. 1

D. K. FEIBACK and E. CROWAN

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Massachusetts Institute of Technology
Department of Mechanical Engineering

D. I. C. 6949

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EXPERIMENTS ON BRITTLE FRACTURE OF STEEL PLATES

1. Modifications of the Griffith theory.

According to the Griffith theory ⁽¹⁾⁽²⁾, the tensile strength of a brittle body containing a surface crack of length ^{*)} c is

$$\sigma = \sqrt{\frac{2 \alpha E}{\pi c}} \quad (1)$$

if the body is a plate thin compared with the length of the crack, and

$$\sigma = \sqrt{\frac{2 \alpha E}{\pi c(1 - \nu^2)}} \quad (2)$$

if it is thick compared with c . In these equations, E is Young's modulus, α the specific surface energy of the surfaces of fracture, and ν Poisson's ratio.

These equations cannot be applied to brittle fracture in normally ductile steels, for the following reason ⁽³⁾. In the Griffith theory, the surface energy α represents the work required for enlarging the crack, per unit area of additional

^{*)} The original calculations of Griffith refer to the two-dimensional case (e.g., a plate containing a crack). The "length" of the crack is then measured on the face of the plate; in the case of an edge crack, it is identical with its depth measured from the edge. The results of the calculation are approximately valid for any surface crack if the radius of curvature of the surface is large compared with the dimensions of the crack and if the width of the crack (measured along the surface) is large compared with its depth. In this case, the depth of the crack corresponds to the crack length in the two-dimensional case and is traditionally called its "length".

surface of fracture. In a completely brittle material, fracture is not accompanied by any plastic deformation; in a low-carbon steel, however, a thin layer adjacent to the surfaces of fracture is plastically distorted, even though the fracture may appear quite brittle: this is recognized from X-ray back reflection photographs⁽⁴⁾. The cause of the distortion is easy to understand: the cleavage planes in neighboring grains do not intersect, in general, at the grain boundary, so that a cleavage in one grain cannot progress smoothly into a neighboring grain. As a rule, cleavage starts independently in several adjacent grains, and the process of separation is then completed by tearing involving plastic deformation at the grain boundaries. In an annealed or hot-rolled steel the X-ray diffraction spots from the individual grains are normally sharp; the intensity of the plastic distortion at the surface of fracture, therefore, can be estimated from the diffuseness of the spots in back-reflection photographs. The effective thickness of the cold-worked layer can be estimated from the rate at which the diffraction spots in successive X-ray photographs become sharper as more and more material is removed by etching from the fracture surface (Figs. 1a, 1b, 1c). Previous estimates of this kind⁽⁴⁾ have led to the order of magnitude of 10^6 ergs/cm² for the plastic distortion work in the brittle fracture surface of a ship steel broken at room temperature; the steel from which the photographs shown in Figs. 1a to 1c were taken gave a probable value of about $2 \cdot 10^6$ ergs/cm².

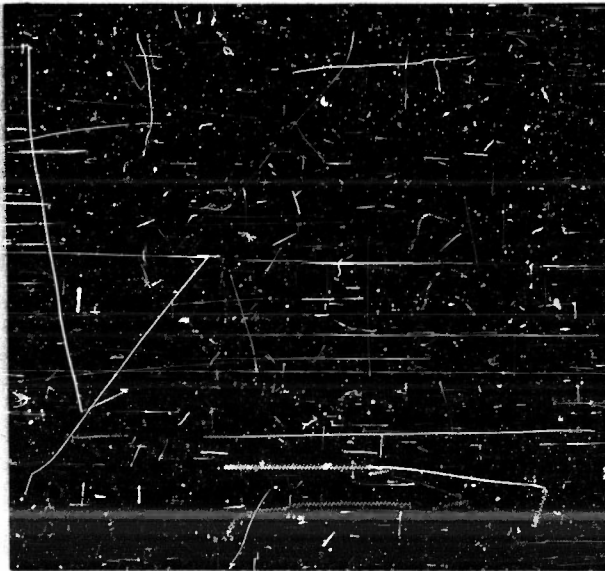


Fig. 1a.
Fracture surface



Fig. 1b: 0.25 mm depth

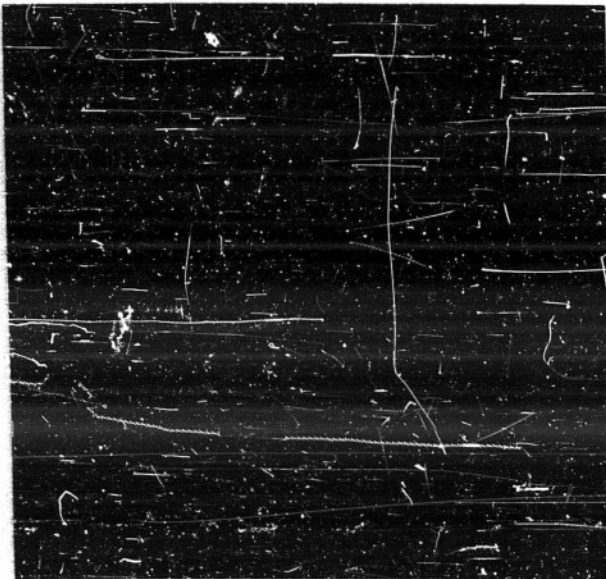


Fig. 1c.
0.58 mm depth

In general, the Griffith theory can be applied only if no plastic deformation occurs during the fracture process. However, if the plastic deformation is confined to a layer at the surface of fracture the thickness of which is small compared with the length of the crack, the plastic work is proportional to the area of the surface of fracture and its

value per unit area can be added to the surface energy, since their sum is the total work required for enlarging the area of the crack wall, by unit amount. The X-ray investigations just mentioned here have shown that the effective thickness of the cold worked layer is of the order of 0.3 or 0.4 mm. This, of course, is far above the usual lengths of the Griffith crack in completely brittle materials like glass; however, brittle fracture in low carbon steel at or around room temperature cannot start without initiation by a crack or a notch far deeper than the thickness of the plastically deformed surface layer. It seems justified, therefore, to assume that the plastic surface work can be taken into account by adding its value p per unit area to the surface energy in the Griffith equation. In this way, eq. (1) becomes

$$\sigma \cong \sqrt{\frac{E(\alpha + p)}{c}} \quad (3)$$

if the factor $\sqrt{2/\pi}$ is omitted. If this equation is satisfied, the elastic energy released during the crack propagation is just sufficient to cover the work $p + \alpha$ required for enlarging the fracture surface of the crack by unit area.

Since, as just mentioned, the plastic surface work p in low carbon steel at room temperature is probably of order of magnitude 10^6 ergs/cm², it is about 1000 times greater than the surface energy which, for hard metals, is between 1000 and 2000 ergs/cm². Consequently, α can be neglected beside p , and eq. (3) written as

$$\sigma \cong \sqrt{\frac{Ep}{c}} \quad (4)$$

The initial purpose of the present investigation was to test experimentally the crack propagation condition eq. (4). The most satisfactory and complete way of doing this would be to measure p independently and to compare the value of σ given by eq. (4) with measurements of the tensile strength of a plate containing an atomically sharp crack of depth c . However, the X-ray measurement of p in the way just mentioned could not be carried out with an accuracy going beyond an order-of-magnitude estimate, for the following reason. In order to establish a relationship between the diffuseness of the X-ray diffraction spots and the amount of plastic strain, a comparison chart of X-ray photographs of the same material for a series of known plastic strains would have to be made. For the present purpose, strains of the order of 1 or 2 per cent are of interest, since X-ray back reflection photographs of the surface of brittle fracture indicate plastic strains of such magnitude. However, plastic strains below the Lüders strain (that present in the Lüders bands before the end of the yield) cannot be produced in macroscopic volumes, so that the required region of the comparison chart would be missing. Furthermore, the distortion in the surface of fracture is very unevenly distributed; some fragments, almost completely torn out of the surface, are severely deformed, while the grains underneath may be only slightly deformed. Owing to this, the type of spottiness seen in the X-ray photographs of the fracture surface (cf. Fig. 1a) is rather different from that obtained with a specimen distorted in tension or compression



Fig. 2.
 $\epsilon = 2 \frac{1}{2} \%$

(see Fig. 3), and the comparison of the two photographs is difficult: the fracture surface reveals a mixture of widely different plastic strains.

After realizing this difficulty, we decided to restrict the investigation to the measurement of fracture stresses of steel plates provided with cracks of different depths, and the comparison of

the observed relationship between σ and c with that in eq. (4). If the measured $\sigma - c$ curve is sufficiently well approximated by eq. (4), it can be used for obtaining a value for the quantity p .

2. Production of specimens containing sharp cracks of given length.

The experiments were carried out on plates from the tanker "Ponaganset" which broke in two in Boston Harbor on December 9, 1947. All specimens discussed in this paper were cut from a $\frac{3}{4}$ in. plate through which the crack ran when the ship broke up; its position in the hull can be identified by means of the report on the "Ponaganset" failure ⁽⁵⁾, where it was designated by the letters "PAD".

The final size of the plate tensile specimens was about 4 in. width by 12 in. length. The width represented the upper limit that

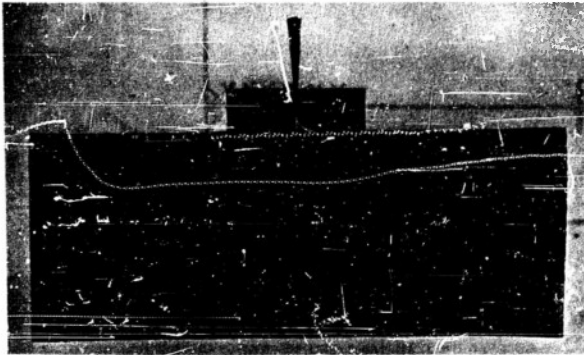


Fig. 3.

could be handled with the largest testing machine at our disposal, a Southwark-Emery hydraulic machine of 300,000 lb capacity. The specimens were provided with an initial brittle crack at the middle of one edge in the following way. First, a specimen with a side flap

according to Fig. 3 was machined, and the flap provided with a notch for inserting a splitting wedge, as shown in the figure. A brittle crack was then produced by hammering the wedge into the notch after the specimen was cooled down by immersion in liquid nitrogen. The length of the crack could be controlled to a certain extent by progressive splitting with a succession of moderate blows with a hammer on the wedge. In addition, any desired length could be achieved accurately when the flap and a certain margin containing the excess length of the crack were trimmed off the plate along a line shown dotted in Fig. 3. The ends of the crack could be recognized clearly by the absence of frost in a narrow zone along the crack. Examination of the fractured specimen showed that the crack front was not straight; in the middle of the plate it was ahead of the ends of the crack visible on the surfaces of the plate.

The presence of a notch on only one side of the plate introduces

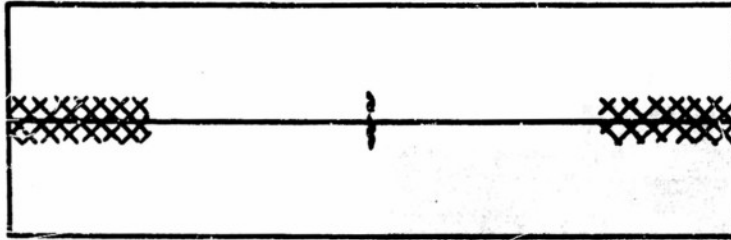


Fig. 4

a certain eccentricity. Although the effect of this can be taken into account approximately⁽⁶⁾, a number of experiments was carried out in which the bending moment due to the asymmetry was compensated in the way shown in Fig. 4. Two strips of 2 in. width each provided with a crack as described above, were tacked together by welding near the ends with the cracks turned towards each other. Any bending due to the eccentricity of the individual strips would force them together and thus would be counteracted by the pressure between them. The experiments described in the following section have shown that the difference between the results obtained with the asymmetrical and the symmetrical (double) specimens was comparatively small.

3. Experiments.

The load was applied to the specimens through wedge grips with roughened inner surfaces; the specimen was gripped so that its free length was about 7 in. with the crack half way between the grips, and loaded at a rate of about 1000 lb/sec. In Table

I the measured tensile strengths are given for all specimens except those which showed disqualifying irregularities of fracture appearance. In most specimens represented in Table I, fracture was completed; in a minority, the crack stopped before the separation was complete. The corresponding data for the symmetrical specimens, cut from the same plate (FAD) as those in Table I, are given separately in Table II.

The relationship between breaking stress and initial crack length is shown in Fig. 5, representing the data of Table I. The curve represents eq. (4), with the value of p that minimizes the sum of the squares of the ordinate difference between the measured points and the curve; this value is $p = 4.9 \times 10^6$ ergs/cm². Thus the experimentally determined specific plastic energy is somewhat higher than the value derived from X-ray observations (see Section (1) above).

The question is whether this discrepancy is a trivial consequence of the inherent inaccuracies of measurement and evaluation, or the representation of a real difference between the plastic surface work estimated from X-ray photographs and the constant p obtained by fitting eq. (4) to the data given in Tables I and II. The latter possibility will be discussed in the following section.

4. Conclusions.

The idea of the experiments described in the preceding section was to provide the specimen before the tensile test with a brittle crack of atomic sharpness and measure the tensile stress required

TABLE I

"Ponaganset" Plate PAD;
specimens with single edge cracks.

Room Temperature Tests
Plate Thickness: 2 cm
Specimen Length: ~ 30 cm

<u>Number</u>	<u>Initial Width:</u> <u>cm.</u>	<u>Initial Crack</u> <u>Length: cm.</u>	<u>Fracture</u> <u>Load: lb</u>	<u>Average Stress</u> <u>kg/cm²</u>
3	10.4	1.7	111,000	2420
4	10.0	0.9	145,000	3300
5	10.0	1.3	134,000	3040
6	10.0	2.6	87,000	1970
8	9.9	2.4	84,000	1930
9	10.0	1.0	137,500	3130
10	10.3	0.5	191,000	4220

TABLE II

"Ponaganset" Plate PAD;
specimens with internal (symmetrical) cracks.

Room Temperature Tests
Plate Thickness: 2 cm
Specimen Length: ~ 30 cm

<u>Number</u>	<u>Initial Width:</u> <u>cm.</u>	<u>Initial Crack</u> <u>Length: cm.</u>	<u>Fracture</u> <u>Load: lb</u>	<u>Average Stress</u> <u>kg/cm²</u>
D-1	10.1	0.68	126,000	2840
D-2	10.1	1.2	106,000	2380
D-3	10.0	1.85	92,000	2090
D-4	10.1	2.0	91,000	2050
D-5	10.1	0.55	154,500	3480

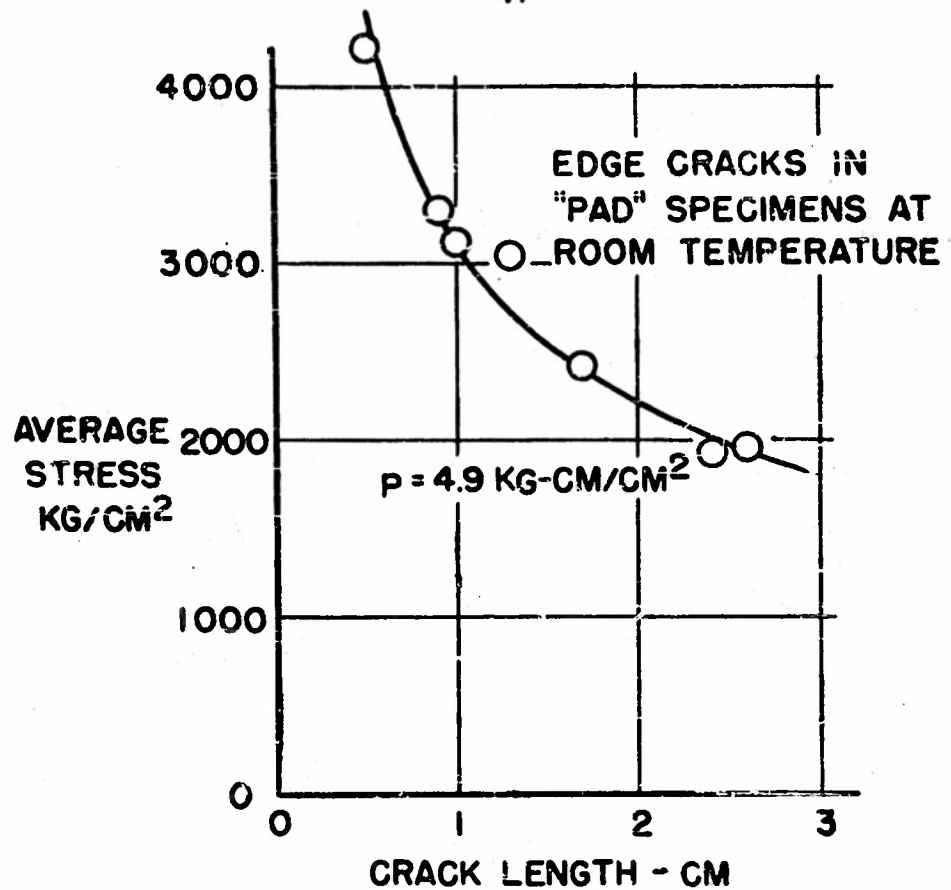


Fig. 5

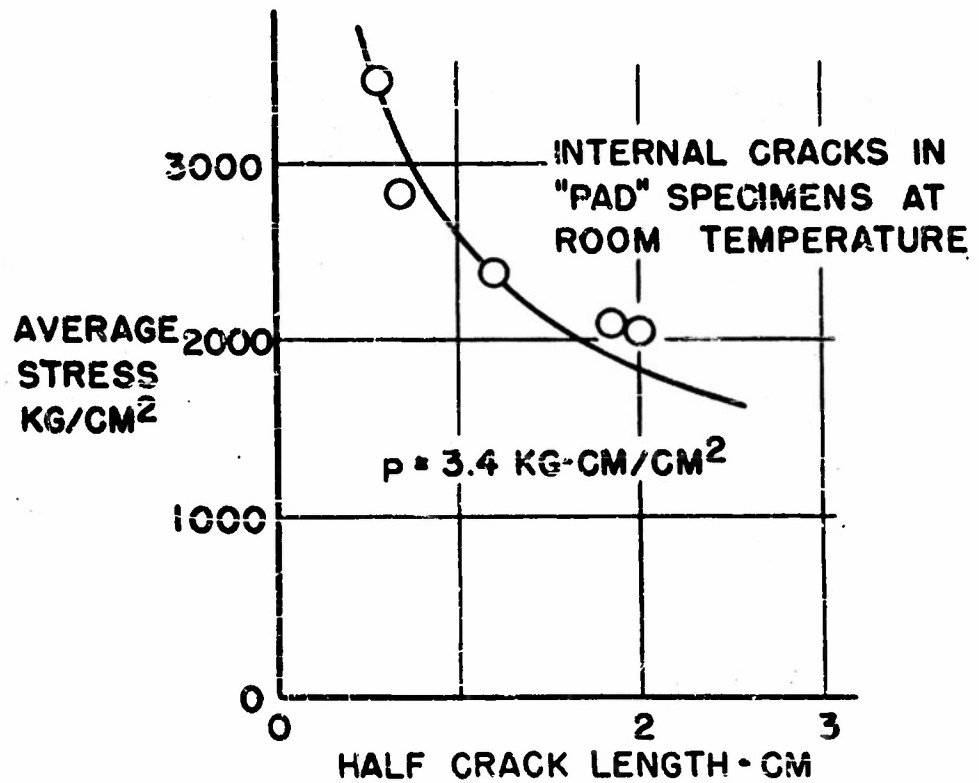


Fig. 6

for starting it to propagate. This, it was hoped, would give a direct experimental test for the crack propagation condition eq. (4).

The plots of the observed fracture stresses against the lengths of the initial cracks, shown in Figs. 5 and 6, seem to provide impressive support for eq. (4): not only do the measured points lie well on curves representing eq. (4), but the value of p derived from them agrees in the order of magnitude with the X-ray estimates. Yet an important feature observable on the fractured specimens casts doubt not only on the significance of this agreement, but also on the foundations of the classical Mesnager-Ludwik triaxial tension theory of notch brittleness. It was easy to recognize at the first visual examination of the fragments that, contrary to the expectation that underlay the program of the investigation, the initial brittle crack never continued to propagate as a brittle crack when the fracture stress was reached. First, considerable plastic deformation took place at the tip of the crack which started to propagate as a fibrous crack. After a very short run, this changed again into a brittle crack which ran across the plate at high velocity (except in the cases discussed further below in which the brittle crack stopped due to load relaxation and had to be restarted by raising the load). This can be seen in Fig. 7b where 1 is the initial brittle crack, 2 the fibrous crack, and 3 the brittle crack developed from the fibrous one in the course of the test. The

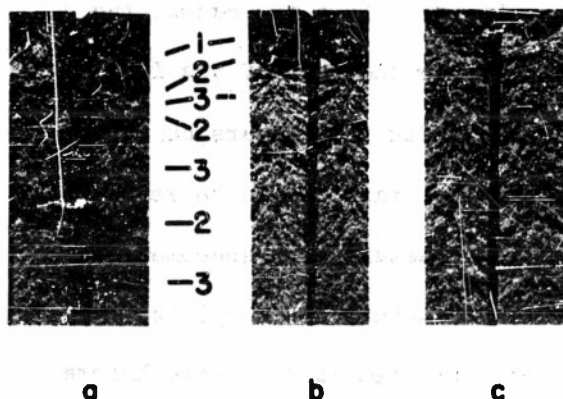


Fig. 7.

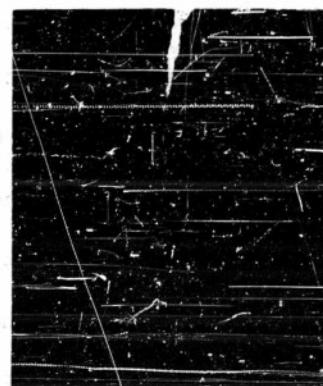


Fig. 8.

contraction of the thickness of the plate by the plastic deformation around the tip of the initiating crack can be recognized in the photograph. Fig. 8 shows the plastic deformation at the tip of the initiating crack as it appears in view upon the face of the plate. Fig. 7c shows the surfaces of fracture in a specimen containing a shorter initiating crack.

Occasionally, when the initial crack was long and therefore the fracture stress low, the second brittle crack did not run through the plate but stopped in it. In such cases, it could be re-started by increasing the load which, owing to the inability of the testing machine to follow the extension of the specimen, had dropped drastically (e. g., to one-half of the initial fracture load) by the time the crack had stopped. However, as in the first instance, the brittle crack did not propagate as such when the load was raised. Again extensive plastic deformation leading to the formation of a narrow zone of fibrous or shear fracture took place

at the tip of the crack, and then the ductile crack changed to a brittle one which ran at high velocity. In a few cases, the crack stopped for a second time owing to the fall of the load, and the whole process of re-starting, with the conversion into a ductile and then again into a brittle crack, could be repeated again. Fig. 7a shows the surface of fracture of a specimen in which the crack stopped twice before fracture was complete.

Why is it that a brittle crack produced in the plate before loading cannot continue to propagate as a brittle crack when the load reaches the necessary value? What are the factors that compel it to change into a ductile crack with copious local plastic deformation before it can change back into a brittle crack? And, in view of this change, do the tensile tests described above have the meaning initially ascribed to them, of giving corresponding values of crack length and crack propagating stress suitable for supporting eq. (4)?

The only simple explanation for the inability of an atomically sharp brittle crack to continue its propagation unchanged when a tensile stress is applied is to attribute it to the absence of velocity when the propagation is resumed under a slowly applied stress. This almost unavoidable conclusion requires a drastic revision of the classical theory of notch brittleness. For the last 50 years, the brittle fracture of ductile steels was attributed, after Mesnager⁽⁷⁾ and Ludwik⁽⁸⁾, to the triaxial tension arising when plastic deformation starts at the tip of a notch or

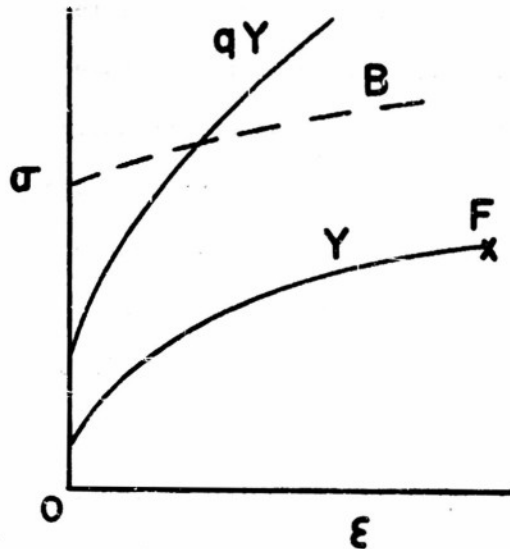


Fig. 9

a crack. Let Y in Fig. 9 be the ordinary uniaxial (true) stress-strain curve of the material in tension; fracture of the ductile (fibrous or shear) type occurs at the point F. Curve B represents the brittle strength (brittle fracture stress); since it lies above the yield stress-strain curve OF, no brittle fracture can occur in the ordinary tensile test. How-

ever, brittle fracture is possible if the maximum tensile stress required for yielding is raised above the values given by the curve OF. This can be done:

- 1) by lowering the temperature: ferrous materials become brittle at a sufficiently low temperature;
- 2) by increasing the rate of straining: ferrous materials have an abnormally high velocity-dependence of the yield stress-strain curve;
- 3) by superposing a hydrostatic tension, e.g., by plastic constraint in a specimen containing a crack or a notch.

The classical theory of notch brittleness saw its main cause in the triaxial tension produced by notch constraint. The way in which this acts can be described very simply without any reference

to triaxiality, in the following manner. In a specimen containing a notch or a crack, plastic deformation starts in the region of stress concentration at the tip of the notch. However, this region is embedded in less highly stressed surroundings that have not reached the point of yielding at the same time; consequently, the region of stress concentration cannot yield freely before the maximum tensile stress in it has increased to the magnitude required for overcoming the constraint of the surroundings in addition to its own resistance to plastic deformation. It can be shown that this effect can raise the maximum tensile stress reached at yielding up to about 3 times the value of the uniaxial tensile yield stress if the notch is very deep and sharp. In the presence of a notch, therefore, the maximum principal tensile stress plotted against the plastic strain will be represented by a curve like that denoted by qY in Fig. 9, and this may intersect the curve B of the brittle strength before the plastic deformation could reach the value necessary for ductile fracture to occur. In this case, the specimen undergoes brittle fracture.

As mentioned above, the Mesnager-Ludwik theory attributed brittle fracture in commonly ductile steels primarily to the development of a triaxial tension during plastic yielding at the tip of a notch or a crack. Such a triaxiality requires the occurrence of plastic deformation; without this, only a small effect due to elastic constraint can be present. How much plastic deformation is necessary for the development of full plastic constraint has never been calculated; it was implicitly assumed that deformation

extending over a very small (perhaps microscopically small) region around the tip of the crack would be sufficient. The observations described above indicate strongly that this expectation was mistaken: it seems that full plastic constraint cannot develop around a crack of macroscopic length without macroscopically observable plastic deformation. Once this is assumed, the explanation of the phenomenon shown in Fig. 7 presents no difficulty. If a fracture appears quite brittle, it cannot involve plastic deformation sufficient for the development of considerable plastic constraint: brittle fracture must then be due to the raising of the yield stress to the level of the brittle strength by the velocity effect. If a crack travels fast, any plastic deformation that occurs at its tip involves an extremely high strain rate and so requires a strongly increased yield stress. With most metals, even the highest velocities cannot increase the yield stress more than 20 or 30 per cent; according to several investigators⁽⁹⁾⁽¹⁰⁾ however, the yield stress (possibly only the upper yield point) of low carbon steels increases by a factor of 2 or 3 at high rates of deformation. This would be quite sufficient to replace fully the highest possible plastic constraint effect. But if the brittleness is due mainly to a velocity effect, the crack put initially into the specimen cannot start propagating in the typical brittle manner under static load because it has no velocity. Consequently, local plastic deformation sets in, and plastic constraint develops until the triaxiality of tension is high enough to change the initial fibrous crack propagation into a brittle one. Once this

has happened, the crack accelerates rapidly and further plastic deformation becomes unnecessary as the velocity effect takes over from the plastic constraint effect the task of raising the maximum tensile stress to the level of the brittle fracture stress.

Fig. 7a shows clearly that, at not too low temperatures, the velocity effect alone may not be sufficient for raising the yield tension to the value of the brittle strength. At the free surface of a plate, no triaxiality can exist; consequently, some plastic deformation with ductile (shear) fracture must occur here if the velocity effect alone is insufficient, and the well known phenomenon of the "shear lip" arises. The plastic deformation produces a slight but sharp constriction running in the form of a shallow and narrow rounded groove along the line of fracture on the surfaces of the plate; this is the same phenomenon as the necking of a tensile specimen, and it is likewise accompanied with the development of transverse tensions in the interior which, added to the velocity effect, lead to brittle fracture everywhere except at the shear lip where, owing to proximity of the free surface, triaxiality becomes too low. Thus, the width of the shear lips gives a measure of how much the velocity effect falls short of being able to produce brittle fracture alone. This is impressively seen in Fig. 7a. When the brittle fracture starts at the fibrous "nail" 2, it soon attains high velocity, and the shear lip is quite narrow. However, as it progresses, the load drops because the testing machine cannot follow the fast expansion of the specimen, and the velocity of the crack decreases. With this, more and more

of the velocity-effect upon the yield stress has to be replaced by a plastic constraint effect, and the width of the shear lip increases until finally the two shear lips join up to a parabolic arc and the crack stops, possibly after it has converted itself into a very narrow margin of a ductile crack.

This picture of brittle fracture changes the classical concept in which triaxiality was the fundamental effect. It seems now that triaxiality alone cannot produce really brittle fracture at all, because, being the result of plastic constraint, it requires considerable plastic deformation. Whenever a fracture is truly brittle (i.e., of extremely low energy consumption), this must be due mainly to the velocity effect, unless the temperature is so low that the relatively small elastic constraint effect is sufficient. Triaxiality is still important in starting off brittle crack propagation, as seen in the above experiments; it is doubtful, however, whether it is indispensable for this purpose. Many service fractures do not reveal any visible trace of plastic deformation at their starting point. This is rarely observed in laboratory experiments, but it occurs sufficiently often under service conditions at completely static loading to exclude doubt about the reality of the phenomenon. Perhaps a cleavage fracture in a single grain, initiated by some insignificant factor, can produce at a sufficiently low temperature a crack velocity high enough for low-energy propagation if the neighboring grains happen to have an orientation favorable for the smooth continuation of the cleavage process.

The observation that the initiating cleavage crack never spread under the applied stress but always changed first into a ductile crack before reverting to the cleavage type raises the question whether the use of the crack propagation condition eq. (4), as it was done in Section (3), has any physical meaning. It is easy to see that the incisive change in the theoretical picture of notch brittleness to which the above experiments have led has clarified the significance of eq. (4). The condition of brittle cleavage fracture now requires a high velocity of the crack, and this can be reached only if the energy to be fed into it (in the form of surface work p) can be obtained from the elastic energy released during its propagation: otherwise, very high velocities of load application would be necessary. Eq. (4) represents the condition that the work needed to enlarge the crack must be supplied from the released elastic energy: it is, therefore, the condition for the high velocity of crack propagation required for brittle cleavage to be reached under static or nearly static loading. From this, however, it does not follow that the fracture stresses observed in the present experiments must have been those given by eq. (4): they could have had any magnitude exceeding this value. That this is a serious possibility is seen at once from the fact that the development of plastic constraint sufficient to provide the required triaxiality necessarily implies a stress condition: it occurs only if the applied stress is large enough to produce a sufficient amount of plastic deformation around the crack. The critical stress required for this purpose

is not directly related to the critical brittle crack propagation stress given by eq. (4); it is likely to be higher, and the fracture stress observed in the experiments described in Sections (2) and (3) must have been the greater of the two critical stresses. Which of the two was greater cannot be recognized from the experiments; however, the circumstance that the values of p derived from the fracture stress measurements were somewhat higher than the value estimated from X-ray photographs would seem to indicate that the triaxiality-producing stress was greater. In this case, of course, the values of p derived from the fracture stresses would have a meaning different from that in eq. (4).

The fact that the measured fracture stresses lie so well on curves representing the crack propagation condition eq. (4) (see Figs. 5 and 6) is no indication that the fracture stresses were identical with the critical crack propagation stress. It can be demonstrated⁽⁶⁾ that the value of the tensile stress at which plastic deformation extends over a region of given size at the tip of the crack is approximately inversely proportional to the square root of the crack length. The measured points, therefore, would lie on a $\sigma - c$ curve of the same mathematical character even if the fracture stress were not the crack propagating stress determined by eq. (4) but the stress required for developing at the tip of the stationary crack the intensity of plastic constraint needed for initiating cleavage fracture.

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